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(54) Title: ULTRA-HIGH STRENGTH AUSAGED STEELS WITH EXCELLENT CRYOGENIC TEMPERATURE TOUGHNESS

(57) Abstract

An ultra-high strength, weldable, low alloy steel with excellent cryogenic temperature toughness in the base plate and in the heat affected zone (HAZ) when welded, having a tensile strength greater than 830 MPa (120 ksi) and a micro-laminate microstructure comprising austenite film layers and fine-grained martensite/lower bainite laths, is prepared by heating a steel slab comprising iron and specified weight percentages of some or all of the additives carbon, manganese, nickel, nitrogen, copper, chromium, molybdenum, silicon, niobium, vanadium, titanium, aluminum, and boron; reducing the slab to form plate in one or more passes in a temperature range in which austenite recrystallizes; finish rolling the plate in one or more passes in a temperature range below the austenite recrystallization temperature and above the Ar3 transformation temperature; quenching the finish rolled plate to a suitable Quench Stop Temperature (QST); stopping the quenching; and either, for a period of time, holding the plate substantially isothermally at the QST or slow-cooling the plate before air cooling, or simply air cooling the plate to ambient temperature.
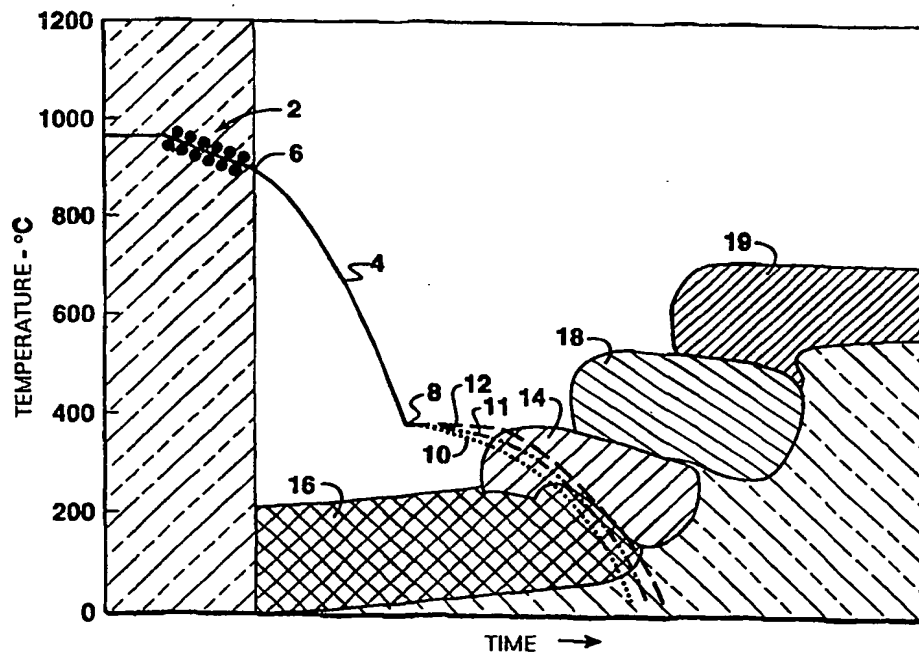
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ULTRA-HIGH STRENGTH AUSAGED STEELS WITH EXCELLENT CRYOGENIC TEMPERATURE TOUGHNESS

5 **FIELD OF THE INVENTION**

This invention relates to ultra-high strength, weldable, low alloy steel plates with excellent cryogenic temperature toughness in both the base plate and in the heat affected zone (HAZ) when welded. Furthermore, this invention relates to a method for producing such steel plates.

10

BACKGROUND OF THE INVENTION

Various terms are defined in the following specification. For convenience, a Glossary of terms is provided herein, immediately preceding the claims.

Frequently, there is a need to store and transport pressurized, volatile fluids at

15 cryogenic temperatures, i.e., at temperatures lower than about -40°C (-40°F). For example, there is a need for containers for storing and transporting pressurized liquefied natural gas (PLNG) at a pressure in the broad range of about 1035 kPa (150 psia) to about 7590 kPa (1100 psia) and at a temperature in the range of about -123°C (-190°F) to about -62°C (-80°F). There is also a need for containers for safely and

20 economically storing and transporting other volatile fluids with high vapor pressure, such as methane, ethane, and propane, at cryogenic temperatures. For such containers to be constructed of a welded steel, the steel must have adequate strength to withstand the fluid pressure and adequate toughness to prevent initiation of a fracture, i.e., a failure event, at the operating conditions, in both the base steel and in the HAZ.

25 The Ductile to Brittle Transition Temperature (DBTT) delineates the two fracture regimes in structural steels. At temperatures below the DBTT, failure in the steel tends to occur by low energy cleavage (brittle) fracture, while at temperatures above the DBTT, failure in the steel tends to occur by high energy ductile fracture. Welded steels used in the construction of storage and transportation containers for the

30 aforementioned cryogenic temperature applications and for other load-bearing, cryogenic temperature service must have DBTTs well below the service temperature in both the base steel and the HAZ to avoid failure by low energy cleavage fracture.

Nickel-containing steels conventionally used for cryogenic temperature structural applications, e.g., steels with nickel contents of greater than about 3 wt%, have low DBTTs, but also have relatively low tensile strengths. Typically, commercially available 3.5 wt% Ni, 5.5 wt% Ni, and 9 wt% Ni steels have DBTTs of

5 about -100°C (-150°F), -155°C (-250°F), and -175°C (-280°F), respectively, and tensile strengths of up to about 485 MPa (70 ksi), 620 MPa (90 ksi), and 830 MPa (120 ksi), respectively. In order to achieve these combinations of strength and toughness, these steels generally undergo costly processing, e.g., double annealing treatment. In the case of cryogenic temperature applications, industry currently uses

10 these commercial nickel-containing steels because of their good toughness at low temperatures, but must design around their relatively low tensile strengths. The designs generally require excessive steel thicknesses for load-bearing, cryogenic temperature applications. Thus, use of these nickel-containing steels in load-bearing, cryogenic temperature applications tends to be expensive due to the high cost of the

15 steel combined with the steel thicknesses required.

On the other hand, several commercially available, state-of-the-art, low and medium carbon high strength, low alloy (HSLA) steels, for example AISI 4320 or 4330 steels, have the potential to offer superior tensile strengths (e.g., greater than about 830 MPa (120 ksi)) and low cost, but suffer from relatively high DBTTs in

20 general and especially in the weld heat affected zone (HAZ). Generally, with these steels there is a tendency for weldability and low temperature toughness to decrease as tensile strength increases. It is for this reason that currently commercially available, state-of-the-art HSLA steels are not generally considered for cryogenic temperature applications. The high DBTT of the HAZ in these steels is generally due

25 to the formation of undesirable microstructures arising from the weld thermal cycles in the coarse grained and intercritically reheated HAZs, i.e., HAZs heated to a temperature of from about the Ac₁ transformation temperature to about the Ac₃ transformation temperature. (See Glossary for definitions of Ac₁ and Ac₃ transformation temperatures.). DBTT increases significantly with increasing grain

30 size and embrittling microstructural constituents, such as martensite-austenite (MA) islands, in the HAZ. For example, the DBTT for the HAZ in a state-of-the-art HSLA steel, X100 linepipe for oil and gas transmission, is higher than about -50°C (-60°F).

There are significant incentives in the energy storage and transportation sectors for the development of new steels that combine the low temperature toughness properties of the above-mentioned commercial nickel-containing steels with the high strength and low cost attributes of the HSLA steels, while also providing excellent weldability and

5   the desired thick section capability, i.e., substantially uniform microstructure and properties (e.g., strength and toughness) in thicknesses greater than about 2.5 cm (1 inch).

In non-cryogenic applications, most commercially available, state-of-the-art, low and medium carbon HSLA steels, due to their relatively low toughness at high

10   strengths, are either designed at a fraction of their strengths or, alternatively, processed to lower strengths for attaining acceptable toughness. In engineering applications, these approaches lead to increased section thickness and therefore, higher component weights and ultimately higher costs than if the high strength potential of the HSLA steels could be fully utilized. In some critical applications,

15   such as high performance gears, steels containing greater than about 3 wt% Ni (such as AISI 48XX, SAE 93XX, etc.) are used to maintain sufficient toughness. This approach leads to substantial cost penalties to access the superior strength of the HSLA steels. An additional problem encountered with use of standard commercial HSLA steels is hydrogen cracking in the HAZ, particularly when low heat input

20   welding is used.

There are significant economic incentives and a definite engineering need for low cost enhancement of toughness at high and ultra-high strengths in low alloy steels. Particularly, there is a need for a reasonably priced steel that has ultra-high strength, e.g., tensile strength greater than 830 MPa (120 ksi), and excellent cryogenic

25   temperature toughness, e.g. DBTT lower than about -73°C (-100°F), both in the base plate and in the HAZ, for use in commercial cryogenic temperature applications.

Consequently, the primary objects of the present invention are to improve the state-of-the-art HSLA steel technology for applicability at cryogenic temperatures in three key areas: (i) lowering of the DBTT to less than about -73°C (-100°F) in the

30   base steel and in the weld HAZ, (ii) achieving tensile strength greater than 830 MPa (120 ksi), and (iii) providing superior weldability. Other objects of the present invention are to achieve the aforementioned HSLA steels with substantially uniform

through-thickness microstructures and properties in thicknesses greater than about 2.5 cm (1 inch) and to do so using current commercially available processing techniques so that use of these steels in commercial cryogenic temperature processes is economically feasible.

5

SUMMARY OF THE INVENTION

Consistent with the above-stated objects of the present invention, a processing methodology is provided wherein a low alloy steel slab of the desired chemistry is reheated to an appropriate temperature then hot rolled to form steel plate and rapidly

10   cooled, at the end of hot rolling, by quenching with a suitable fluid, such as water, to a suitable Quench Stop Temperature (QST) to produce a micro-laminate microstructure comprising, preferably, about 2 vol% to about 10 vol% austenite film layers and about 90 vol% to about 98 vol% laths of predominantly fine-grained martensite and fine-grained lower bainite. In one embodiment of this invention, the steel plate is then

15   air cooled to ambient temperature. In another embodiment, the steel plate is held substantially isothermally at the QST for up to about five (5) minutes, followed by air cooling to ambient temperature. In yet another embodiment, the steel plate is slow-cooled at a rate lower than about 1.0°C per second (1.8°F/sec) for up to about five (5) minutes, followed by air cooling to ambient temperature. As used in describing the

20   present invention, quenching refers to accelerated cooling by any means whereby a fluid selected for its tendency to increase the cooling rate of the steel is utilized, as opposed to air cooling the steel to ambient temperature.

Also, consistent with the above-stated objects of the present invention, steels processed according to the present invention are especially suitable for many

25   cryogenic temperature applications in that the steels have the following characteristics, preferably for steel plate thicknesses of about 2.5 cm (1 inch) and greater: (i) DBTT lower than about -73°C (-100°F) in the base steel and in the weld HAZ, (ii) tensile strength greater than 830 MPa (120 ksi), preferably greater than about 860 MPa (125 ksi), and more preferably greater than about 900 MPa (130 ksi),

30   (iii) superior weldability, (iv) substantially uniform through-thickness microstructure and properties, and (v) improved toughness over standard, commercially available, HSLA steels. These steels can have a tensile strength of greater than about 930 MPa

(135 ksi), or greater than about 965 MPa (140 ksi), or greater than about 1000 MPa (145 ksi).

DESCRIPTION OF THE DRAWINGS

5   The advantages of the present invention will be better understood by referring to the following detailed description and the attached drawings in which:

FIG. 1 is a schematic continuous cooling transformation (CCT) diagram showing how the ausaging process of the present invention produces micro-laminate microstructure in a steel according to the present invention;

10   FIG. 2A (Prior Art) is a schematic illustration showing a cleavage crack propagating through lath boundaries in a mixed microstructure of lower bainite and martensite in a conventional steel;

FIG. 2B is a schematic illustration showing a tortuous crack path due to the presence of the austenite phase in the micro-laminate microstructure in a steel according

15   to the present invention;

FIG. 3A is a schematic illustration of austenite grain size in a steel slab after reheating according to the present invention;

FIG. 3B is a schematic illustration of prior austenite grain size (see Glossary) in a steel slab after hot rolling in the temperature range in which austenite recrystallizes, but

20   prior to hot rolling in the temperature range in which austenite does not recrystallize, according to the present invention; and

FIG. 3C is a schematic illustration of the elongated, pancake grain structure in austenite, with very fine effective grain size in the through-thickness direction, of a steel plate upon completion of TMCP according to the present invention.

25   While the present invention will be described in connection with its preferred embodiments, it will be understood that the invention is not limited thereto. On the contrary, the invention is intended to cover all alternatives, modifications, and equivalents which may be included within the spirit and scope of the invention, as defined by the appended claims.

30

DETAILED DESCRIPTION OF THE INVENTION

The present invention relates to the development of new HSLA steels meeting the above-described challenges. The invention is based on a novel combination of steel chemistry and processing for providing both intrinsic and microstructural

5 toughening to lower DBTT as well as to enhance toughness at high tensile strengths. Intrinsic toughening is achieved by the judicious balance of critical alloying elements in the steel, as described in detail in this specification. Microstructural toughening results from achieving a very fine effective grain size as well as promoting micro-laminate microstructure. Referring to FIG. 2B, the micro-laminate

10 microstructure of steels according to this invention is preferably comprised of alternating laths 28, of predominantly either fine-grained lower bainite or fine-grained martensite, and austenite film layers 30. Preferably, the average thickness of the austenite film layers 30 is less than about 10% of the average thickness of the laths 28. Even more preferably, the average thickness of the austenite film layers 30 is

15 about 10 nm and the average thickness of the laths 28 is about 0.2 microns.

Ausaging is used in the present invention to facilitate formation of the micro-laminate microstructure by promoting retention of the desired austenite film layers at ambient temperatures. As is familiar to those skilled in the art, ausaging is a process wherein aging of austenite in a heated steel takes place prior to the steel

20 cooling to the temperature range where austenite typically transforms to bainite and/or martensite. It is known in the art that ausaging promotes thermal stabilization of austenite. The unique steel chemistry and processing combination of this invention provides for a sufficient delay time in the start of the bainite transformation after quenching is stopped to allow for adequate aging of the austenite for formation of the

25 austenite film layers in the micro-laminate microstructure. For example, referring now to FIG. 1, a steel processed according to this invention undergoes controlled rolling 2 within the temperature ranges indicated (as described in greater detail hereinafter); then the steel undergoes quenching 4 from the start quench point 6 until the stop quench point (i.e., QST) 8. After quenching is stopped at the stop quench

30 point (QST) 8, (i) in one embodiment, the steel plate is held substantially isothermally at the QST for a period of time, preferably up to about 5 minutes, and then air cooled to ambient temperature, as illustrated by the dashed line 12, (ii) in another

embodiment, the steel plate is slow cooled from the QST at a rate lower than about

1.0°C per second (1.8°F/sec) for up to about 5 minutes, prior to allowing the steel

plate to air cool to ambient temperature, as illustrated by the dash-dot-dot line 11, (iii)

in still another embodiment, the steel plate may be allowed to air cool to ambient

temperature, as illustrated by the dotted line 10. In any of the embodiments, austenite

film layers are retained after formation of lower bainite laths in the lower bainite

region 14 and martensite laths in the martensite region 16. The upper bainite region

18 and ferrite/pearlite region 19 are avoided. In the steels of the present invention,

enhanced ausaging occurs due to the novel combination of steel chemistry and

processing described in this specification.

The bainite and martensite constituents and the austenite phase of the

micro-laminate microstructure are designed to exploit the superior strength attributes

of fine-grained lower bainite and fine-grained lath martensite, and the superior

cleavage fracture resistance of austenite. The micro-laminate microstructure is

optimized to substantially maximize tortuosity in the crack path, thereby enhancing

the crack propagation resistance to provide significant microstructural toughening.

In accordance with the foregoing, a method is provided for preparing an

ultra-high strength, steel plate having a micro-laminate microstructure comprising

about 2 vol% to about 10 vol% austenite film layers and about 90 vol% to about 98

vol% laths of predominantly fine-grained martensite and fine-grained lower bainite,

said method comprising the steps of: (a) heating a steel slab to a reheating

temperature sufficiently high to (i) substantially homogenize the steel slab, (ii)

dissolve substantially all carbides and carbonitrides of niobium and vanadium in the

steel slab, and (iii) establish fine initial austenite grains in the steel slab; (b) reducing

the steel slab to form steel plate in one or more hot rolling passes in a first

temperature range in which austenite recrystallizes; (c) further reducing the steel plate

in one or more hot rolling passes in a second temperature range below about the T_{nr}

temperature and above about the Ar_3 transformation temperature; (d) quenching the

steel plate at a cooling rate of about 10°C per second to about 40°C per second

(18°F/sec - 72°F/sec) to a Quench Stop Temperature (QST) below about the M_s

transformation temperature plus 100°C (180°F) and above about the M_s

transformation temperature; and (e) stopping said quenching. In one embodiment, the method of this invention further comprises the step of allowing the steel plate to air cool to ambient temperature from the QST. In another embodiment, the method of this invention further comprises the step of holding the steel plate substantially

5   isothermally at the QST for up to about 5 minutes prior to allowing the steel plate to air cool to ambient temperature. In yet another embodiment, the method of this invention further comprises the step of slow-cooling the steel plate from the QST at a rate lower than about 1.0°C per second (1.8°F/sec) for up to about 5 minutes prior to allowing the steel plate to air cool to ambient temperature. This processing facilitates

10 transformation of the microstructure of the steel plate to about 2 vol% to about 10 vol% of austenite film layers and about 90 vol% to about 98 vol% laths of predominantly fine-grained martensite and fine-grained lower bainite. (See Glossary for definitions of T_{nr} temperature, and of Ar_3 and M_s transformation temperatures.)

To ensure ambient and cryogenic temperature toughness, the laths in the

15 micro-laminate microstructure preferably comprise predominantly lower bainite or martensite. It is preferable to substantially minimize the formation of embrittling constituents such as upper bainite, twinned martensite and MA. As used in describing the present invention, and in the claims, "predominantly" means at least about 50 volume percent. The remainder of the microstructure can comprise additional

20 fine-grained lower bainite, additional fine-grained lath martensite, or ferrite. More preferably, the microstructure comprises at least about 60 volume percent to about 80 volume percent lower bainite or lath martensite. Even more preferably, the microstructure comprises at least about 90 volume percent lower bainite or lath martensite.

25 A steel slab processed according to this invention is manufactured in a customary fashion and, in one embodiment, comprises iron and the following alloying elements, preferably in the weight ranges indicated in the following Table I:

Table I

Alloying Element	Range (wt%)
carbon (C)	0.04 - 0.12, more preferably 0.04 - 0.07
manganese (Mn)	0.5 - 2.5, more preferably 1.0 - 1.8
nickel (Ni)	1.0 - 3.0, more preferably 1.5 - 2.5
copper (Cu)	0.1 - 1.0, more preferably 0.2 - 0.5
molybdenum (Mo)	0.1 - 0.8, more preferably 0.2 - 0.4
niobium (Nb)	0.02 - 0.1, more preferably 0.02 - 0.05
titanium (Ti)	0.008 - 0.03, more preferably 0.01 - 0.02
aluminum (Al)	0.001 - 0.05, more preferably 0.005 - 0.03
nitrogen (N)	0.002 - 0.005, more preferably 0.002 - 0.003

Chromium (Cr) is sometimes added to the steel, preferably up to about 1.0 wt%, and more preferably about 0.2 wt% to about 0.6 wt%.

Silicon (Si) is sometimes added to the steel, preferably up to about 0.5 wt%, more preferably about 0.01 wt% to about 0.5 wt%, and even more preferably about 0.05 wt% to about 0.1 wt%.

The steel preferably contains at least about 1 wt% nickel. Nickel content of the steel can be increased above about 3 wt% if desired to enhance performance after welding. Each 1 wt% addition of nickel is expected to lower the DBTT of the steel by about 10°C (18°F). Nickel content is preferably less than 9 wt%, more preferably less than about 6 wt%. Nickel content is preferably minimized in order to minimize cost of the steel. If nickel content is increased above about 3 wt%, manganese content can be decreased below about 0.5 wt% down to 0.0 wt%.

Boron (B) is sometimes added to the steel, preferably up to about 0.0020 wt%, and more preferably about 0.0006 wt% to about 0.0010 wt%.

Additionally, residuals are preferably substantially minimized in the steel. Phosphorous (P) content is preferably less than about 0.01 wt%. Sulfur (S) content is preferably less than about 0.004 wt%. Oxygen (O) content is preferably less than about 0.002 wt%.

Processing of the Steel Slab

(1) Lowering of DBTT

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Achieving a low DBTT, e.g., lower than about -73°C (-100°F), is a key challenge in the development of new HSLA steels for cryogenic temperature applications. The technical challenge is to maintain/increase the strength in the present HSLA technology while lowering the DBTT, especially in the HAZ. The present invention utilizes a combination of alloying and processing to alter both the intrinsic as well as microstructural contributions to fracture resistance in a way to produce a low alloy steel with excellent cryogenic temperature properties in the base plate and in the HAZ, as hereinafter described.

In this invention, microstructural toughening is exploited for lowering the base steel DBTT. This microstructural toughening consists of refining prior austenite grain size, modifying the grain morphology through thermo-mechanical controlled rolling processing (TMCP), and producing a micro-laminate microstructure within the fine grains, all aimed at enhancing the interfacial area of the high angle boundaries per unit volume in the steel plate. As is familiar to those skilled in the art, "grain" as used herein means an individual crystal in a polycrystalline material, and "grain boundary" as used herein means a narrow zone in a metal corresponding to the transition from one crystallographic orientation to another, thus separating one grain from another. As used herein, a "high angle grain boundary" is a grain boundary that separates two adjacent grains whose crystallographic orientations differ by more than about 8°. Also, as used herein, a "high angle boundary or interface" is a boundary or interface that effectively behaves as a high angle grain boundary, i.e., tends to deflect a propagating crack or fracture and, thus, induces tortuosity in a fracture path.

The contribution from TMCP to the total interfacial area of the high angle boundaries per unit volume, Sv, is defined by the following equation:

30

$$Sv = \frac{1}{d}\left(1 + R + \frac{1}{R}\right) + 0.63(r - 30)$$

where:

    d is the average austenite grain size in a hot-rolled steel plate prior to rolling in the temperature range in which austenite does not recrystallize (prior austenite grain size);

5

    R is the reduction ratio (original steel slab thickness/final steel plate thickness); and

10

    r is the percent reduction in thickness of the steel due to hot rolling in the temperature range in which austenite does not recrystallize.

It is well known in the art that as the Sv of a steel increases, the DBTT decreases, due to crack deflection and the attendant tortuosity in the fracture path at

15   the high angle boundaries. In commercial TMCP practice, the value of R is fixed for a given plate thickness and the upper limit for the value of r is typically 75. Given fixed values for R and r, Sv can only be substantially increased by decreasing d, as evident from the above equation. To decrease d in steels according to the present invention, Ti-Nb microalloying is used in combination with optimized TMCP

20   practice. For the same total amount of reduction during hot rolling/deformation, a steel with an initially finer average austenite grain size will result in a finer finished average austenite grain size. Therefore, in this invention the amount of Ti-Nb additions are optimized for low reheating practice while producing the desired austenite grain growth inhibition during TMCP. Referring to FIG. 3A, a relatively

25   low reheating temperature, preferably between about 955°C and about 1065°C (1750°F - 1950°F), is used to obtain initially an average austenite grain size D' of less than about 120 microns in reheated steel slab 32' before hot deformation. Processing according to this invention avoids the excessive austenite grain growth that results from the use of higher reheating temperatures, i.e., greater than about 1095°C

30   (2000°F), in conventional TMCP. To promote dynamic recrystallization induced grain refining, heavy per pass reductions greater than about 10% are employed during hot rolling in the temperature range in which austenite recrystallizes. Referring now to FIG. 3B, processing according to this invention provides an average prior austenite grain size D" (i.e., d) of less than about 30 microns, preferably less than about 20

35   microns, and even more preferably less than about 10 microns, in steel slab 32" after

hot rolling (deformation) in the temperature range in which austenite recrystallizes, but prior to hot rolling in the temperature range in which austenite does not recrystallize. Additionally, to produce an effective grain size reduction in the through-thickness direction, heavy reductions, preferably exceeding about 70%

5   cumulative, are carried out in the temperature range below about the T_{nr} temperature but above about the Ar_3 transformation temperature. Referring now to FIG. 3C, TMCP according to this invention leads to the formation of an elongated, pancake structure in austenite in a finish rolled steel plate 32''' with very fine effective grain size D''' in the through-thickness direction, e.g., effective grain size D''' less than about

10 10 microns, preferably less than about 8 microns, and even more preferably less than about 5 microns, thus enhancing the interfacial area of high angle boundaries, e.g. 33, per unit volume in steel plate 32''', as will be understood by those skilled in the art.

In somewhat greater detail, a steel according to this invention is prepared by forming a slab of the desired composition as described herein; heating the slab to a

15 temperature of from about 955°C to about 1065°C (1750°F - 1950°F); hot rolling the slab to form steel plate in one or more passes providing about 30 percent to about 70 percent reduction in a first temperature range in which austenite recrystallizes, i.e., above about the T_{nr} temperature, and further hot rolling the steel plate in one or more passes providing about 40 percent to about 80 percent reduction in a second

20 temperature range below about the T_{nr} temperature and above about the Ar_3 transformation temperature. The hot rolled steel plate is then quenched at a cooling rate of about 10°C per second to about 40°C per second (18°F/sec - 72°F/sec) to a suitable QST below about the M_s transformation temperature plus 100°C (180°F) and above about the M_s transformation temperature, at which time the quenching is

25 terminated. In one embodiment of this invention, after quenching is terminated the steel plate is allowed to air cool to ambient temperature from the QST, as illustrated by the dotted line 10 of FIG. 1. In another embodiment of this invention, after quenching is terminated the steel plate is held substantially isothermally at the QST for a period of time, preferably up to about 5 minutes, and then air cooled to ambient

30 temperature, as illustrated by the dashed line 12 of FIG. 1. In yet another embodiment

as illustrated by the dash-dot-dot line 11 of FIG. 1, the steel plate is slow-cooled from the QST at a rate slower than that of air cooling, i.e., at a rate lower than about 1°C per second (1.8°F/sec), preferably for up to about 5 minutes. In at least one embodiment of this invention, the M_s transformation temperature is about 350°C

5   (662°F) and, therefore, the M_s transformation temperature plus 100°C (180°F) is

about 450°C (842°F).

The steel plate may be held substantially isothermally at the QST by any suitable means, as are known to those skilled in the art, such as by placing a thermal blanket over the steel plate. The steel plate may be slow-cooled after quenching is

10 terminated by any suitable means, as are known to those skilled in the art, such as by placing an insulating blanket over the steel plate.

As is understood by those skilled in the art, as used herein percent reduction in thickness refers to percent reduction in the thickness of the steel slab or plate prior to the reduction referenced. For purposes of explanation only, without thereby limiting this

15 invention, a steel slab of about 25.4 cm (10 inches) thickness may be reduced about 50% (a 50 percent reduction), in a first temperature range, to a thickness of about 12.7 cm (5 inches) then reduced about 80% (an 80 percent reduction), in a second temperature range, to a thickness of about 2.5 cm (1 inch). As used herein, "slab" means a piece of steel having any dimensions.

20 The steel slab is preferably heated by a suitable means for raising the temperature of substantially the entire slab, preferably the entire slab, to the desired reheating temperature, e.g., by placing the slab in a furnace for a period of time. The specific reheating temperature that should be used for any steel composition within the range of the present invention may be readily determined by a person skilled in the art, either by

25 experiment or by calculation using suitable models. Additionally, the furnace temperature and reheating time necessary to raise the temperature of substantially the entire slab, preferably the entire slab, to the desired reheating temperature may be readily determined by a person skilled in the art by reference to standard industry publications.

Except for the reheating temperature, which applies to substantially the entire

30 slab, subsequent temperatures referenced in describing the processing method of this invention are temperatures measured at the surface of the steel. The surface

temperature of steel can be measured by use of an optical pyrometer, for example, or by any other device suitable for measuring the surface temperature of steel. The cooling rates referred to herein are those at the center, or substantially at the center, of the plate thickness; and the Quench Stop Temperature (QST) is the highest, or

5   substantially the highest, temperature reached at the surface of the plate, after quenching is stopped, because of heat transmitted from the mid-thickness of the plate. For example, during processing of experimental heats of a steel composition according to this invention, a thermocouple is placed at the center, or substantially at the center, of the steel plate thickness for center temperature measurement, while the

10   surface temperature is measured by use of an optical pyrometer. A correlation between center temperature and surface temperature is developed for use during subsequent processing of the same, or substantially the same, steel composition, such that center temperature may be determined via direct measurement of surface temperature. Also, the required temperature and flow rate of the quenching fluid to

15   accomplish the desired accelerated cooling rate may be determined by one skilled in the art by reference to standard industry publications.

For any steel composition within the range of the present invention, the temperature that defines the boundary between the recrystallization range and non-recrystallization range, the T_{nr} temperature, depends on the chemistry of the steel,

20   particularly the carbon concentration and the niobium concentration, on the reheating temperature before rolling, and on the amount of reduction given in the rolling passes. Persons skilled in the art may determine this temperature for a particular steel according to this invention either by experiment or by model calculation. Similarly, the Ar_3 and M_s transformation temperatures referenced herein may be determined by persons skilled

25   in the art for any steel according to this invention either by experiment or by model calculation.

The TMCP practice thus described leads to a high value of Sv. Additionally, referring again to FIG. 2B, the micro-laminate microstructure produced during ausaging further increases the interfacial area by providing numerous high angle

30   interfaces 29 between the laths 28 of predominantly lower bainite or martensite and the austenite film layers 30. This micro-laminate configuration, as schematically

illustrated in FIG. 2B, may be compared to the conventional bainite/martensite lath

structure without the interlath austenite film layers, as illustrated in FIG. 2A. The

conventional structure schematically illustrated in FIG. 2A is characterized by low

angle boundaries 20 (i.e., boundaries that effectively behave as low angle grain

5   boundaries (see Glossary)), e.g., between laths 22 of predominantly lower bainite and

martensite; and thus, once a cleavage crack 24 is initiated, it can propagate through

the lath boundaries 20 with little change in direction. In contrast, the micro-laminate

microstructure in the steels of the current invention, as illustrated by FIG. 2B, leads to

significant tortuosity in the crack path. This is because a crack 26 that is initiated in a

10   lath 28, e.g., of lower bainite or martensite, for instance, will tend to change planes,

i.e., change directions, at each high angle interface 29 with austenite film layers 30

due to the different orientation of cleavage and slip planes in the bainite and

martensite constituents and the austenite phase. Additionally, the austenite film layers

30 provide blunting of an advancing crack 26 resulting in further energy absorption

15   before the crack 26 propagates through the austenite film layers 30. The blunting

occurs for several reasons. First, the FCC (as defined herein) austenite does not

exhibit DBTT behavior and shear processes remain the only crack extension

mechanism. Secondly, when the load/strain exceeds a certain higher value at the

crack tip, the metastable austenite can undergo a stress or strain induced

20   transformation to martensite leading to TRansformation Induced Plasticity (TRIP).

TRIP can lead to significant energy absorption and lower the crack tip stress intensity.

Finally, the lath martensite that forms from TRIP processes will have a different

orientation of the cleavage and slip plane than that of the pre-existing bainite or lath

martensite constituents making the crack path more tortuous. As illustrated by FIG.

25   2B, the net result is that the crack propagation resistance is significantly enhanced in

the micro-laminate microstructure.

       The bainite/austenite or martensite/austenite interfaces of steels according to

the present invention have excellent interfacial bond strengths and this forces crack

deflection rather than interfacial debonding. The fine-grained lath martensite and

30   fine-grained lower bainite occur as packets with high angle boundaries between the

packets. Several packets are formed within a pancake. This provides a further degree

of structural refinement leading to enhanced tortuosity for crack propagation through

these packets within the pancake. This leads to substantial increase in Sv and consequently, lowering of DBTT.

Although the microstructural approaches described above are useful for lowering DBTT in the base steel plate, they are not fully effective for maintaining

5   sufficiently low DBTT in the coarse grained regions of the weld HAZ. Thus, the present invention provides a method for maintaining sufficiently low DBTT in the coarse grained regions of the weld HAZ by utilizing intrinsic effects of alloying elements, as described in the following.

Leading ferritic cryogenic temperature steels are generally based on

10   body-centered cubic (BCC) crystal lattice. While this crystal system offers the potential for providing high strengths at low cost, it suffers from a steep transition from ductile to brittle fracture behavior as the temperature is lowered. This can be fundamentally attributed to the strong sensitivity of the critical resolved shear stress (CRSS) (defined herein) to temperature in BCC systems, wherein CRSS rises steeply

15   with a decrease in temperature thereby making the shear processes and consequently ductile fracture more difficult. On the other hand, the critical stress for brittle fracture processes such as cleavage is less sensitive to temperature. Therefore, as the temperature is lowered, cleavage becomes the favored fracture mode, leading to the onset of low energy brittle fracture. The CRSS is an intrinsic property of the steel and

20   is sensitive to the ease with which dislocations can cross slip upon deformation; that is, a steel in which cross slip is easier will also have a low CRSS and hence a low DBTT. Some face-centered cubic (FCC) stabilizers such as Ni are known to promote cross slip, whereas BCC stabilizing alloying elements such as Si, Al, Mo, Nb and V discourage cross slip. In the present invention, content of FCC stabilizing alloying

25   elements, such as Ni and Cu, is preferably optimized, taking into account cost considerations and the beneficial effect for lowering DBTT, with Ni alloying of preferably at least about 1.0 wt% and more preferably at least about 1.5 wt%; and the content of BCC stabilizing alloying elements in the steel is substantially minimized.

As a result of the intrinsic and microstructural toughening that results from the

30   unique combination of chemistry and processing for steels according to this invention, the steels have excellent cryogenic temperature toughness in both the base plate and the HAZ after welding. DBTTs in both the base plate and the HAZ after welding of

these steels are lower than about -73°C (-100°F) and can be lower than about -107°C (-160°F).

(2) Tensile Strength greater than 830 MPa (120 ksi) and Through-Thickness
5   Uniformity of Microstructure and Properties

The strength of micro-laminate structure is primarily determined by the carbon content of the lath martensite and lower bainite. In the low alloy steels of the present invention, ausaging is carried out to produce austenite content in the steel plate of
10   preferably about 2 volume percent to about 10 volume percent, more preferably at least about 5 volume percent. Ni and Mn additions of about 1.0 wt% to about 3.0 wt% and of about 0.5 wt% to about 2.5 wt%, respectively, are especially preferred for providing the desired volume fraction of austenite and the delay in bainite start for ausaging. Copper additions of preferably about 0.1 wt% to about 1.0 wt% also
15   contribute to the stabilization of austenite during ausaging.

In the present invention, the desired strength is obtained at a relatively low carbon content with the attendant advantages in weldability and excellent toughness in both the base steel and in the HAZ. A minimum of about 0.04 wt% C is preferred in the overall alloy for attaining tensile strength greater than 830 MPa (120 ksi).

20   While alloying elements, other than C, in steels according to this invention are substantially inconsequential as regards the maximum attainable strength in the steel, these elements are desirable to provide the required through-thickness uniformity of microstructure and strength for plate thickness greater than about 2.5 cm (1 inch) and for a range of cooling rates desired for processing flexibility. This is important as the
25   actual cooling rate at the mid section of a thick plate is lower than that at the surface. The microstructure of the surface and center can thus be quite different unless the steel is designed to eliminate its sensitivity to the difference in cooling rate between the surface and the center of the plate. In this regard, Mn and Mo alloying additions, and especially the combined additions of Mo and B, are particularly effective. In the
30   present invention, these additions are optimized for hardenability, weldability, low DBTT and cost considerations. As stated previously in this specification, from the point of view of lowering DBTT, it is essential that the total BCC alloying additions

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be kept to a minimum. The preferred chemistry targets and ranges are set to meet
these and the other requirements of this invention.

(3) Superior Weldability For Low Heat Input Welding

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The steels of this invention are designed for superior weldability. The most
important concern, especially with low heat input welding, is cold cracking or
hydrogen cracking in the coarse grained HAZ. It has been found that for steels of the
present invention, cold cracking susceptibility is critically affected by the carbon
10   content and the type of HAZ microstructure, not by the hardness and carbon
equivalent, which have been considered to be the critical parameters in the art. In
order to avoid cold cracking when the steel is to be welded under no or low preheat
(lower than about 100°C (212°F)) welding conditions, the preferred upper limit for
carbon addition is about 0.1 wt%. As used herein, without limiting this invention in
15   any aspect, "low heat input welding" means welding with arc energies of up to about
2.5 kilojoules per millimeter (kJ/mm) (7.6 kJ/inch).

Lower bainite or auto-tempered lath martensite microstructures offer superior
resistance to cold cracking. Other alloying elements in the steels of this invention are
carefully balanced, commensurate with the hardenability and strength requirements,
20   to ensure the formation of these desirable microstructures in the coarse grained HAZ.

Role of Alloying Elements in the Steel Slab

The role of the various alloying elements and the preferred limits on their
25   concentrations for the present invention are given below:

Carbon (C) is one of the most effective strengthening elements in steel. It also
combines with the strong carbide formers in the steel such as Ti, Nb, and V to provide
grain growth inhibition and precipitation strengthening. Carbon also enhances
hardenability, i.e., the ability to form harder and stronger microstructures in the steel
30   during cooling. If the carbon content is less than about 0.04 wt%, it is generally not
sufficient to induce the desired strengthening, viz., greater than 830 MPa (120 ksi)
tensile strength, in the steel. If the carbon content is greater than about 0.12 wt%,

generally the steel is susceptible to cold cracking during welding and the toughness is reduced in the steel plate and its HAZ on welding. Carbon content in the range of about 0.04 wt% to about 0.12 wt% is preferred to produce the desired HAZ microstructures, viz., auto-tempered lath martensite and lower bainite. Even more

5   preferably, the upper limit for carbon content is about 0.07 wt%.

Manganese (Mn) is a matrix strengthener in steels and also contributes strongly to the hardenability. Mn addition is useful for obtaining the desired bainite transformation delay time needed for ausaging. A minimum amount of 0.5 wt% Mn is preferred for achieving the desired high strength in plate thickness exceeding about

10 2.5 cm (1 inch), and a minimum of at least about 1.0 wt% Mn is even more preferred. However, too much Mn can be harmful to toughness, so an upper limit of about 2.5 wt% Mn is preferred in the present invention. This upper limit is also preferred to substantially minimize centerline segregation that tends to occur in high Mn and continuously cast steels and the attendant through-thickness non-uniformity in

15 microstructure and properties. More preferably, the upper limit for Mn content is about 1.8 wt%. If nickel content is increased above about 3 wt%, the desired high strength can be achieved without the addition of manganese. Therefore, in a broad sense, up to about 2.5 wt% manganese is preferred.

Silicon (Si) is added to steel for deoxidation purposes and a minimum of about

20 0.01 wt% is preferred for this purpose. However, Si is a strong BCC stabilizer and thus raises DBTT and also has an adverse effect on the toughness. For these reasons, when Si is added, an upper limit of about 0.5 wt% Si is preferred. More preferably, the upper limit for Si content is about 0.1 wt%. Silicon is not always necessary for deoxidation since aluminum or titanium can perform the same function.

25 Niobium (Nb) is added to promote grain refinement of the rolled microstructure of the steel, which improves both the strength and toughness. Niobium carbide precipitation during hot rolling serves to retard recrystallization and to inhibit grain growth, thereby providing a means of austenite grain refinement. For these reasons, at least about 0.02 wt% Nb is preferred. However, Nb is a strong BCC

30 stabilizer and thus raises DBTT. Too much Nb can be harmful to the weldability and HAZ toughness, so a maximum of about 0.1 wt% is preferred. More preferably, the upper limit for Nb content is about 0.05 wt%.

Titanium (Ti), when added in a small amount, is effective in forming fine titanium nitride (TiN) particles which refine the grain size in both the rolled structure and the HAZ of the steel. Thus, the toughness of the steel is improved. Ti is added in such an amount that the weight ratio of Ti/N is preferably about 3.4. Ti is a strong

5   BCC stabilizer and thus raises DBTT. Excessive Ti tends to deteriorate the toughness of the steel by forming coarser TiN or titanium carbide (TiC) particles. A Ti content below about 0.008 wt% generally can not provide sufficiently fine grain size or tie up the N in the steel as TiN while more than about 0.03 wt% can cause deterioration in toughness. More preferably, the steel contains at least about 0.01 wt% Ti and no

10 more than about 0.02 wt% Ti.

Aluminum (Al) is added to the steels of this invention for the purpose of deoxidation. At least about 0.001 wt% Al is preferred for this purpose, and at least about 0.005 wt% Al is even more preferred. Al ties up nitrogen dissolved in the HAZ. However, Al is a strong BCC stabilizer and thus raises DBTT. If the Al

15 content is too high, i.e., above about 0.05 wt%, there is a tendency to form aluminum oxide (Al_2O_3) type inclusions, which tend to be harmful to the toughness of the steel and its HAZ. Even more preferably, the upper limit for Al content is about 0.03 wt%.

Molybdenum (Mo) increases the hardenability of steel on direct quenching, especially in combination with boron and niobium. Mo is also desirable for

20 promoting ausaging. For these reasons, at least about 0.1 wt% Mo is preferred, and at least about 0.2 wt% Mo is even more preferred. However, Mo is a strong BCC stabilizer and thus raises DBTT. Excessive Mo helps to cause cold cracking on welding, and also tends to deteriorate the toughness of the steel and HAZ, so a maximum of about 0.8 wt% Mo is preferred, and a maximum of about 0.4 wt% Mo is

25 even more preferred.

Chromium (Cr) tends to increase the hardenability of steel on direct quenching. In small additions, Cr leads to stabilization of austenite. Cr also improves corrosion resistance and hydrogen induced cracking (HIC) resistance. Similar to Mo, excessive Cr tends to cause cold cracking in weldments, and tends to deteriorate the

30 toughness of the steel and its HAZ, so when Cr is added a maximum of about 1.0 wt% Cr is preferred. More preferably, when Cr is added the Cr content is about 0.2 wt% to about 0.6 wt%.

Nickel (Ni) is an important alloying addition to the steels of the present
invention to obtain the desired DBTT, especially in the HAZ. It is one of the
strongest FCC stabilizers in steel. Ni addition to the steel enhances the cross slip and
thereby lowers DBTT. Although not to the same degree as Mn and Mo additions, Ni

5   addition to the steel also promotes hardenability and therefore through-thickness
uniformity in microstructure and properties, such as strength and toughness, in thick
sections. Ni addition is also useful for obtaining the desired bainite transformation
delay time needed for ausaging. For achieving the desired DBTT in the weld HAZ,
the minimum Ni content is preferably about 1.0 wt%, more preferably about 1.5 wt%.

10   Since Ni is an expensive alloying element, the Ni content of the steel is preferably less
than about 3.0 wt%, more preferably less than about 2.5 wt%, more preferably less
than about 2.0 wt%, and even more preferably less than about 1.8 wt%, to
substantially minimize cost of the steel.

Copper (Cu) is a desirable alloying addition to stabilize austenite to produce

15   the micro-laminate microstructure. Preferably at least about 0.1 wt%, more preferably
at least about 0.2 wt%, of Cu is added for this purpose. Cu is also an FCC stabilizer
in steel and can contribute to lowering of DBTT in small amounts. Cu is also
beneficial for corrosion and HIC resistance. At higher amounts, Cu induces excessive
precipitation hardening via ε-copper precipitates. This precipitation, if not properly

20   controlled, can lower the toughness and raise the DBTT both in the base plate and
HAZ. Higher Cu can also cause embrittlement during slab casting and hot rolling,
requiring co-additions of Ni for mitigation. For the above reasons, an upper limit of
about 1.0 wt% Cu is preferred, and an upper limit of about 0.5 wt% is even more
preferred.

25   Boron (B) in small quantities can greatly increase the hardenability of steel
and promote the formation of steel microstructures of lath martensite, lower bainite,
and ferrite by suppressing the formation of upper bainite, both in the base plate and
the coarse grained HAZ. Generally, at least about 0.0004 wt% B is needed for this
purpose. When boron is added to steels of this invention, from about 0.0006 wt% to

30   about 0.0020 wt% is preferred, and an upper limit of about 0.0010 wt% is even more
preferred. However, boron may not be a required addition if other alloying in the
steel provides adequate hardenability and the desired microstructure.

(4) Preferred Steel Composition When Post Weld Heat Treatment (PWHT) Is Required

5   PWHT is normally carried out at high temperatures, e.g., greater than about 540°C (1000°F). The thermal exposure from PWHT can lead to a loss of strength in the base plate as well as in the weld HAZ due to softening of the microstructure associated with the recovery of substructure (i.e., loss of processing benefits) and coarsening of cementite particles. To overcome this, the base steel chemistry as

10   described above is preferably modified by adding a small amount of vanadium. Vanadium is added to give precipitation strengthening by forming fine vanadium carbide (VC) particles in the base steel and HAZ upon PWHT. This strengthening is designed to offset substantially the strength loss upon PWHT. However, excessive VC strengthening is to be avoided as it can degrade the toughness and raise DBTT

15   both in the base plate and its HAZ. In the present invention an upper limit of about 0.1 wt% is preferred for V for these reasons. The lower limit is preferably about 0.02 wt%. More preferably, about 0.03 wt% to about 0.05 wt% V is added to the steel.

This step-out combination of properties in the steels of the present invention provides a low cost enabling technology for certain cryogenic temperature operations,

20   for example, storage and transport of natural gas at low temperatures. These new steels can provide significant material cost savings for cryogenic temperature applications over the current state-of-the-art commercial steels, which generally require far higher nickel contents (up to about 9 wt%) and are of much lower strengths (less than about 830 MPa (120 ksi)). Chemistry and microstructure design

25   are used to lower DBTT and provide uniform mechanical properties in the through-thickness for section thicknesses exceeding about 2.5 cm. (1 inch). These new steels preferably have nickel contents lower than about 3 wt%, tensile strength greater than 830 MPa (120 ksi), preferably greater than about 860 MPa (125 ksi), and more preferably greater than about 900 MPa (130 ksi), ductile to brittle transition

30   temperatures (DBTTs) below about -73°C (-100°F), and offer excellent toughness at DBTT. These new steels can have a tensile strength of greater than about 930 MPa (135 ksi), or greater than about 965 MPa (140 ksi), or greater than about 1000 MPa (145 ksi). Nickel content of these steel can be increased above about 3 wt% if

desired to enhance performance after welding. Each 1 wt% addition of nickel is expected to lower the DBTT of the steel by about 10°C (18°F). Nickel content is preferably less than 9 wt%, more preferably less than about 6 wt%. Nickel content is preferably minimized in order to minimize cost of the steel.

5   While the foregoing invention has been described in terms of one or more preferred embodiments, it should be understood that other modifications may be made without departing from the scope of the invention, which is set forth in the following claims.

Glossary of terms:

Ac$_1$ transformation temperature: the temperature at which austenite begins to form during heating;

5   Ac$_3$ transformation temperature: the temperature at which transformation of ferrite to austenite is completed during heating;

Al$_2$O$_3$:   aluminum oxide;

Ar$_3$ transformation temperature: the temperature at which austenite begins to transform to ferrite during cooling;

10   BCC: body-centered cubic;

cooling rate: cooling rate at the center, or substantially at the center, of the plate thickness;

CRSS (critical resolved shear stress): an intrinsic property of a steel, sensitive to the ease with which dislocations can cross slip upon

15   deformation, that is, a steel in which cross slip is easier will also have a low CRSS and hence a low DBTT;

cryogenic temperature: any temperature lower than about -40°C (-40°F);

DBTT (Ductile to Brittle

20   Transition Temperature):   delineates the two fracture regimes in structural steels; at temperatures below the DBTT, failure tends to occur by low energy cleavage (brittle) fracture, while at temperatures above the DBTT, failure tends to occur by high energy ductile

25   fracture;

FCC:	face-centered cubic;
grain:	an individual crystal in a polycrystalline material;
grain boundary:	a narrow zone in a metal corresponding to the transition from one crystallographic orientation to another, thus separating one grain from another;
HAZ:	heat affected zone;
HIC:	hydrogen induced cracking;
high angle boundary or interface:	boundary or interface that effectively behaves as a high angle grain boundary, i.e., tends to deflect a propagating crack or fracture and, thus, induces tortuosity in a fracture path;
high angle grain boundary:	a grain boundary that separates two adjacent grains whose crystallographic orientations differ by more than about 8°;
HSLA:	high strength, low alloy;
intercritically reheated:	heated (or reheated) to a temperature of from about the Ac_1 transformation temperature to about the Ac_3 transformation temperature;
low alloy steel:	a steel containing iron and less than about 10 wt% total alloy additives;

low angle grain boundary:	a grain boundary that separates two adjacent grains whose crystallographic orientations differ by less than about 8°;
low heat input welding:	welding with arc energies of up to about 2.5 kJ/mm (7.6 kJ/inch);
MA:	martensite-austenite;
M_s transformation temperature:	the temperature at which transformation of austenite to martensite starts during cooling;
predominantly:	as used in describing the present invention, means at least about 50 volume percent;
prior austenite grain size:	average austenite grain size in a hot-rolled steel plate prior to rolling in the temperature range in which austenite does not recrystallize;
quenching:	as used in describing the present invention, accelerated cooling by any means whereby a fluid selected for its tendency to increase the cooling rate of the steel is utilized, as opposed to air cooling;

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Quench Stop Temperature (QST):   the highest, or substantially the highest, temperature reached at the surface of the plate, after quenching is stopped, because of heat transmitted from the mid-thickness of the plate;

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slab: a piece of steel having any dimensions;

Sv :   total interfacial area of the high angle boundaries per unit volume in steel plate;

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tensile strength: in tensile testing, the ratio of maximum load to original cross-sectional area;

15   TiC: titanium carbide;

TiN: titanium nitride;

T_{nr} temperature: the temperature below which austenite does not recrystallize; and

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TMCP: thermo-mechanical controlled rolling processing.

We Claim:

1.   A method for preparing a steel plate having a micro-laminate microstructure
comprising about 2 vol% to about 10 vol% of austenite film layers and about
90 vol% to about 98 vol% laths of predominantly fine-grained martensite and
fine-grained lower bainite, said method comprising the steps of:

(a)   heating a steel slab to a reheating temperature sufficiently high to (i)
substantially homogenize said steel slab, (ii) dissolve substantially all
carbides and carbonitrides of niobium and vanadium in said steel slab,
and (iii) establish fine initial austenite grains in said steel slab;

(b)   reducing said steel slab to form steel plate in one or more hot rolling
passes in a first temperature range in which austenite recrystallizes;

(c)   further reducing said steel plate in one or more hot rolling passes in a
second temperature range below about the T_{nr} temperature and above
about the Ar_3 transformation temperature;

(d)   quenching said steel plate at a cooling rate of about 10°C per second to
about 40°C per second (18°F/sec - 72°F/sec) to a Quench Stop
Temperature below about the M_s transformation temperature plus
100°C (180°C) and above about the M_s transformation temperature;
and

(e)   stopping said quenching, so as to facilitate transformation of said steel
plate to a micro-laminate microstructure of about 2 vol% to about 10
vol% of austenite film layers and about 90 vol% to about 98 vol% laths
of predominantly fine-grained martensite and fine-grained lower
bainite.

2.   The method of claim 1 wherein said reheating temperature of step (a) is
between about 955°C and about 1065°C (1750°F - 1950°F).

5   3.   The method of claim 1 wherein said fine initial austenite grains of step (a)
have a grain size of less than about 120 microns.

4.   The method of claim 1 wherein a reduction in thickness of said steel slab of
about 30% to about 70% occurs in step (b).

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5.   The method of claim 1 wherein a reduction in thickness of said steel plate of
about 40% to about 80% occurs in step (c).

6.   The method of claim 1 further comprising the step of allowing said steel plate
15   to air cool to ambient temperature from said Quench Stop Temperature.

7.   The method of claim 1 further comprising the step of holding said steel plate
substantially isothermally at said Quench Stop Temperature for up to about 5
minutes.

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8.   The method of claim 1 further comprising the step of slow-cooling said steel
plate at said Quench Stop Temperature at a rate lower than about 1.0°C per
second (1.8°F/sec) for up to about 5 minutes.

9.   The method of claim 1 wherein said steel slab of step (a) comprises iron and

the following alloying elements in the weight percents indicated:

about 0.04% to about 0.12% C,

at least about 1% Ni,

about 0.1% to about 1.0% Cu,

about 0.1% to about 0.8% Mo,

about 0.02% to about 0.1% Nb,

about 0.008% to about 0.03% Ti,

about 0.001% to about 0.05% Al, and

about 0.002% to about 0.005% N.

10.   The method of claim 9 wherein said steel slab comprises less than about 6

wt% Ni.

11.   The method of claim 9 wherein said steel slab comprises less than about 3

wt% Ni and additionally comprises about 0.5 wt% to about 2.5 wt% Mn.

12.   The method of claim 9 wherein said steel slab further comprises at least one

additive selected from the group consisting of (i) up to about 1.0 wt% Cr, (ii) up

to about 0.5 wt% Si, (iii) about 0.02 wt% to about 0.10 wt% V, and (iv) up to

about 2.5 wt% Mn.

13.   The method of claim 9 wherein said steel slab further comprises about 0.0004

wt% to about 0.0020 wt% B.

14.   The method of claim 1 wherein, after step (e), said steel plate has a DBTT

lower than about -73°C(-100°F) in both said base plate and its HAZ and has a

tensile strength greater than 830 MPa (120 ksi).

15. A steel plate having a micro-laminate microstructure comprising about 2 vol% to about 10 vol% of austenite film layers and about 90 vol% to about 98 vol% laths of fine-grained martensite and fine-grained lower bainite, having a tensile strength greater than 830 MPa (120 ksi), and having a DBTT of lower than about -73°C (-100°F) in both said steel plate and its HAZ, and wherein said steel plate is produced from a reheated steel slab comprising iron and the following alloying elements in the weight percents indicated:

about 0.04% to about 0.12% C,

at least about 1% Ni,

about 0.1% to about 1.0% Cu,

about 0.1% to about 0.8% Mo,

about 0.02% to about 0.1% Nb,

about 0.008% to about 0.03% Ti,

about 0.001% to about 0.05% Al, and

about 0.002% to about 0.005% N.

16. The steel plate of claim 15 wherein said steel slab comprises less than about 6 wt% Ni.

17. The steel plate of claim 15 wherein said steel slab comprises less than about 3 wt% Ni and additionally comprises about 0.5 wt% to about 2.5 wt% Mn.

18. The steel plate of claim 15 further comprising at least one additive selected from the group consisting of (i) up to about 1.0 wt% Cr, (ii) up to about 0.5 wt% Si, (iii) about 0.02 wt% to about 0.10 wt% V, and (iv) up to about 2.5 wt% Mn.

19. The steel plate of claim 15 further comprising about 0.0004 wt% to about 0.0020 wt% B.

20. The steel plate of claim 15, wherein said micro-laminate microstructure is optimized to substantially maximize crack path tortuosity by thermo-mechanical controlled rolling processing that provides a plurality of high angle interfaces between said laths of fine-grained martensite and fine-grained lower bainite and said austenite film layers.

21. A method for enhancing the crack propagation resistance of a steel plate, said method comprising processing said steel plate to produce a micro-laminate microstructure comprising about 2 vol% to about 10 vol% of austenite film layers and about 90 vol% to about 98 vol% laths of predominantly fine-grained martensite and fine-grained lower bainite, said micro-laminate microstructure being optimized to substantially maximize crack path tortuosity by thermo-mechanical controlled rolling processing that provides a plurality of high angle interfaces between said laths of fine-grained martensite and fine-grained lower bainite and said austenite film layers.

22. The method of claim 21 wherein said crack propagation resistance of said steel plate is further enhanced, and crack propagation resistance of the HAZ of said steel plate when welded is enhanced, by adding at least about 1.0 wt% Ni and at least about 0.1 wt% Cu, and by substantially minimizing addition of BCC stabilizing elements.
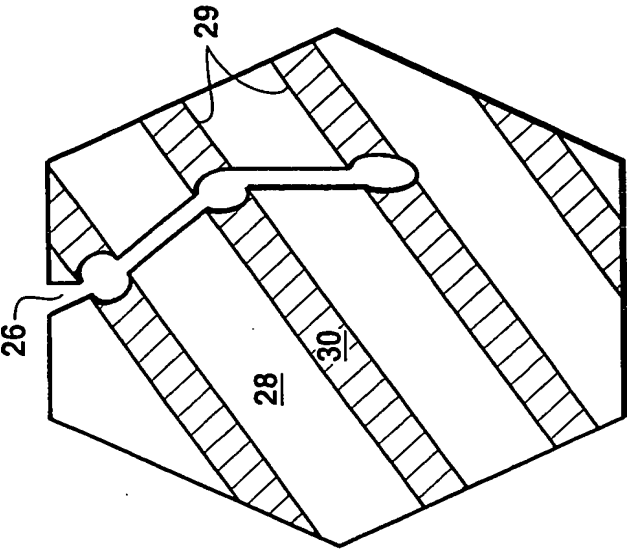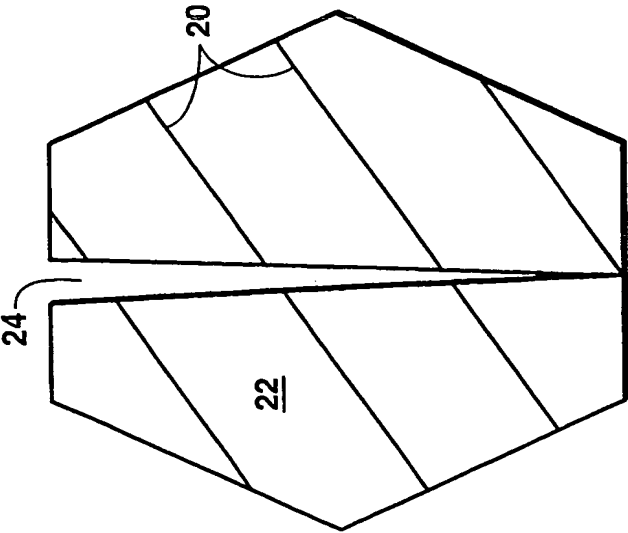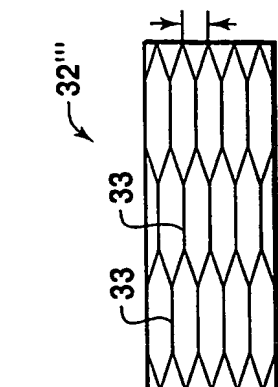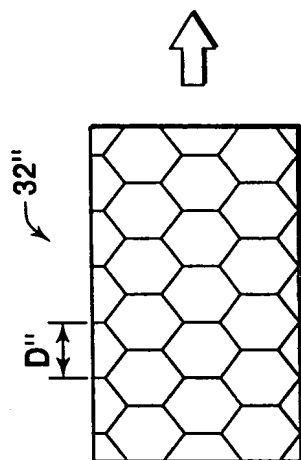
FIG. 1

FIG. 2B

FIG. 2A
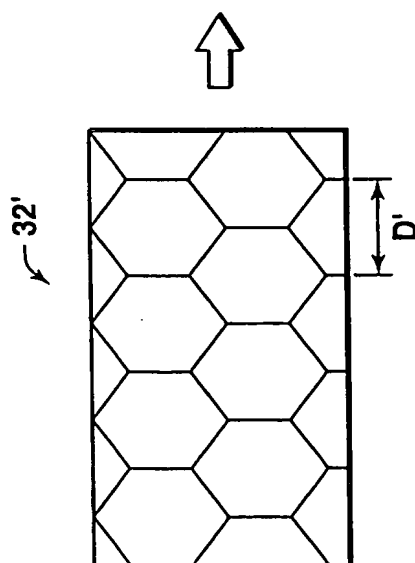(PRIOR ART)

FIG. 3A

FIG. 3B

FIG. 3C

INTERNATIONAL SEARCH REPORT

International application No.
PCT/US98/12705

A. CLASSIFICATION OF SUBJECT MATTER

IPC(6) :C21D 08/00

US CL :148/654, 336; 420/119

According to International Patent Classification (IPC) or to both national classification and IPC

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)

U.S. : 148/654, 336; 420/119

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)

CAS, WPIDS

search terms: iron, Fe, copper, Cu, nickel, Ni, martensite

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
Y	JP 63-062843 A (KOBE STEEL, LTD) 19 March 1988 (19-03-88), abstract.	1-22
Y	JP 09-235617 A (SUMITOMO METAL INDUS., LTD) 09 September 1997 (09-09-97), abstract.	1-22
Y	WO 9623083 A (NIPPON STEEL CORP.) 01 August 1996 (01-08-96), abstract.	1-22
Y	JP 59-013055 A (SUMQ) 23 January 1984 (23-01-84), abstract.	1-22
Y	US 4,878,955 A (HOSHINO ET AL) 07 November 1989 (07-11-89), whole document.	1-22
Y	US 5,183,198 A (TAMEHIRO ET AL) 02 February 1993 (02-02-93), whole document.	1-22

[X] Further documents are listed in the continuation of Box C.	[ ] See patent family annex.

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Date of the actual completion of the international search	Date of mailing of the international search report
07 OCTOBER 1998	**26 OCT 1998**

Name and mailing address of the ISA/US Commissioner of Patents and Trademarks Box PCT Washington, D.C. 20231 Facsimile No. (703) 305-3230	Authorized officer SIKYIN IP Telephone No. (703) 308-0651

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BNSDOCID: <WO_____9932670A1_I_>

International application No.

PCT/US98/12705

C (Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
Y,E	US 5,785,924 A (BEGUINOT ET AL) 28 July 1998 (28-07-98), whole document.	1-22

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